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TECHNICAL REPORT ARSCD-TR-83016

INTRODUCTION TO COMPOSITE MATERIALS

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AUGUST 1984



U.S. ARMY ARMAMENT RESEARCH AND DEVELOPMENT CENTER

FIRE CONTROL AND SMALL CALIBER WEAPON SYSTEMS LABORATORY

DOVER, NEW JERSEY

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20. ABSTRACT (Continue on reverse side if necessary and identify by block number) Advanced composite materials are light, stiff, and extremely strong, but suffer poor transverse strength and high price. Additional requirements for strength at elevated temperatures and stability under severe environmental conditions are satisfied by increasing use of metal matrix materials. Two ways to increase the composite strength is by the use of larger diameter filaments and by heat treating aluminum alloy matrix. (cont)		

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Block 20. Abstract (cont)

Advanced composites are used mainly as panels or structural members in space vehicles or aircraft; a lot of new development work is required before composites can be used for machined parts.

Subjects covered in this report are general information on reinforcements, interface and bonding, micromechanics, consolidation processes, mechanical properties of composite materials, improved mechanical properties, and applications.

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CONTENTS

	Page
Introduction	1
Reinforcement	2
Interface and Bonding	2
Types of Interface	2
Bonding in Composites	3
Mechanical Aspects of the Interface	4
Micromechanics	
Introduction	5
Density of Composites	5
Composite Stresses and Strains	6
Rule of Mixture	7
Strength	8
Consolidation Processes	9
Diffusion Bonding	10
Plasma Spray Bonding	11
Electroforming	11
Liquid Metal Infiltration	11
High Energy Rate Forming	12
Hot Rolling Bonding	12
Mechanical Properties of Composite Materials	12
Strength	12
Tensile Properties	16
Compression	19
Impact	19
Elevated Temperature Tensile Strength	20
Improved Mechanical Properties	21
Introduction	21
Improved Filaments	21
Improved Composites	21
Applications	
Introduction	23
Examples of Applications	24
Cost	24
Further Applications	24

	Page
References	27
Distribution List	49

TABLES

	Page
1 Filament properties	29
2 Boron type filament properties	29
3 Typical fiber properties	29
4 Typical strengths of unidirectional composites in MPa	30
5 Strength parameters in stress space for unidirectional composites	30
6 Strength parameters in strain space for unidirectional composites (dimensionless)	30
7 Tensile properties of boron/aluminum and boron/titanium composites	31
8 Properties of 50 v/o boron/aluminum	31
9 Properties of 50 v/o Borsic/Ti	31
10 Axial tensile strength of 5.6 mil B/Al	32
11 Axial tensile strength of 5.7 mil Borsic/Al composites	33
12 Transverse tensile properties of 5.6 mil B/Al	34
13 Transverse tensile properties of 5.7 mil Borsic/Al	35
14 Axial tensile strength of 48 v/o 5.6 mil B/6061 Al Composites	36

Figures

	Page
1 Schematic of principal bond types	37
2 Typical stress-strain relations of fiber, matrix and composite	37
3 Composite materials fabrication process	37
4 A schematic diagram of the winding apparatus	38
5 Schematic diagram of the apparatus for vacuum casting	39
6 Composite billet for high energy rate forming composite plates	40
7 Hot-roll bonding	41
8 On-axis positive and negative shears	41
9 Uniaxial longitudinal tensile and compressive tests	41
10 Typical stress-strain curves for 50 Vol% unidirectional boron/aluminum composites tested parallel and perpendicular to the filament	42
11 Schematic stress-strain curve for filamentary reinforced metals showing three regions	42
12 Typical compressive stress-strain curves for 50 Vol% unidirectional boron/aluminum composites tested parallel and perpendicular (90°) to the filament	43
13 Variation of longitudinal and transverse compressive strength with boron content for unidirectional boron/aluminum composites	44
14 Variation of impact energy with filament content for various notch-filament configurations for 46 Vol% Borsic-6061 aluminum unidirectional composites	44
15 Variation of longitudinal (0°) tensile strength with test temperature for unidirectional boron (Borsic)-aluminum composites	45
16 Variation of longitudinal (0°) tensile strength with test temperature for various cross - and angle-ply boron-aluminum composites	45

	Page
17 Variation of transverse (90°) tensile strength with test temperature for unidirectional boron-aluminum composites	46
18 A typical histogram for the strength of boron, SiC coated boron and B ₄ C/boron filament	46
19 Metal matrix composite cost history and projections	47
20 Current and projected costs for fibers and composites	47

INTRODUCTION

The development of composite materials has been a subject of intensive interest for at least 25 years. However, the concept of using two or more elemental materials combined to form the constituent phases of a composite solid has been employed ever since materials were first used.

From the earliest uses, the goals for composite development have been to achieve a combination of properties not achievable by any of the elemental materials acting alone; thus a solid mix could be prepared from constituents which, by themselves, could not satisfy a particular design requirement.

Historically, the ancient Egyptians made use of laminated wood, and the Romans used plywood for fine furniture. In the 15th century, German armor showed a typical laminated structure of alternating layers of steel and iron. The fine Japanese blade combines several different steels or steels and irons to provide an extremely hard and keen edge along with a softer body. In the 17th century Indian flintrock, different kinds of iron and steel were first combined into strip, contorted and twisted into a helix, and then welded together to form the gun tube.

Today, composite engineering materials are employed in ever increasing volume and in increasingly diverse fields, because:

- o They combine the properties of their component parts to obtain composite properties which may be new or unique.
- o They make it easier or less costly to obtain certain properties than is possible with solid materials.
- o Special physical, chemical, electrical and magnetic properties might be involved, thereby exciting interest from specialists in various disciplines.

The basic principles - orientation of structure and strength properties, combination of hardness, toughness, lightness, strength, durability and other engineering attributes - are essentially the same with modern composite engineering.

A great interest in mechanics of heterogeneous systems arose in the engineering and scientific community during the last quarter of a century. Demands on materials imposed by today's advanced technologies have become so diverse and severe that they often cannot be met by simple single-component material acting alone. It is frequently necessary to combine several materials into a composite to which each constituent not only contributes its share, but whose combined action transcends the sum of the individual properties and provides new performance unattainable by the constituents acting alone. Space vehicle, heat shields, rocket propellants, buildings and many others impose requirements that are best met and in many instances met only, by composite materials (ref 1).

REINFORCEMENT

For some years now, a wide selection of high modulus, high strength, low density, and often refractory filamentary materials has become available as candidate reinforcements for metallic materials. A number of these are listed with typical mechanical properties in Table 1 (ref 2) with the more familiar glass and metal filaments given for comparison. For metals, the filaments of major interest have been boron, silicon-carbide, alumina, refractory wires, and graphite. In many cases, the preparative methods result in problems such as defects and residual stresses, which mitigate against maximizing the mechanical properties of composite structure. A silicon-carbide coated boron with slightly lower tensile strength than boron (Borsic) is also made and used where boron-metal interactions degrade the filament.

Boron filaments are produced by chemical vapor deposition of boron on hot 0.5 mil tungsten wire substrate from boron trichloride and hydrogen at approximately 1000°C. The typical filament is 4 mils in diameter with strengths averaging 450×10^3 psi. A filamentary fatigue life in excess of a million cycles (using a tension-zero-tension cycle at 150 cycle/min) has been measured using a cyclic load of half the mean tensile strength. The density of 2.6 g/cm^3 is slightly greater than E-glass, but the specific modulus of boron is far superior to that of glass fiber.

During the deposition process, the tungsten wire is at least partially converted to tungsten diboride and cooled rapidly from the hot zone in high-speed processing. Residual stresses are generated in the boron filament from the differences in thermal expansion between the boron deposit and substrate. These residual stresses make the filaments susceptible to longitudinal cracking.

The most encouraging recent advance in reinforcement for metal matrix composites is undoubtedly the commercial development of a wide diameter boron filament. This results in major improvements in the properties of metal composite structures such as boron/aluminum (B/Al) and boron/titanium (B/Ti) in the transverse direction to the axis of reinforcement. It considerably reduces the cost of the filament. The wide diameter filament is easier to handle, reportedly breaks and splits far less and has a better surface consistency than the thinner filament.

The properties of those filaments currently commercially available are given in Table 2 (ref 3).

INTERFACE AND BONDING

Types of Interface

One of the first systematic examinations of types of interface was made by Petrasek and Weeton (1964) (ref 4) who extended the earlier work of Jech et al (1960) (ref 4) on copper-tungsten by study of tungsten-reinforced copper alloys. Three interface types were noted with these alloy matrices, although interpretation of the results was made somewhat difficult by the effects of the alloying elements on the tungsten wire. The types are: those where recrystallization occurred at the periphery of the wire, those where a new phase formed at the

interface, and those where a mutual solution occurred between the matrix and filament.

A general scheme for the classification of interfaces has been developed and is based on the type of chemical reaction occurring between the filament and matrix.

Class I, filament and matrix mutually nonreactive and insoluble.

Class II, filament and matrix mutually nonreactive but soluble.

Class III, filament and matrix react⁽¹⁾ to form compound(s) at interface.

Clear cut definitions between the classes are not always possible, but the groupings provide a systematic background against which to discuss their characteristics.

Bonding in Composites

Six types of bonds are commonly found. These are: the mechanical bond, the dissolution and wetting bond, the oxide bond, the reaction bond, the exchange bond, and mixed bonds. Figure 1 (ref 4) presents schematic examples of some of the principal bond types.

Mechanical Bonding

It requires an absence of any chemical source of bonding from Van der Waals forces, and involves mechanical interlocking. It can arise from mechanical interlocking or from frictional effects arising from the contraction of the matrix on the filament. However, the absence of any chemical source of bond will cause a composite to be very weak under transverse loads and this bond is not believed to be useful in composite technology.

Dissolution and Wetting Bond

A contact angle of less than 90° occurs in wetting and is also characteristic of dissolution. If wetting is assumed to be accompanied by some dissolution, however small, then this bonding characteristic covers both extremes of mutual solubility. Elimination of adsorbed gases and of contaminant films must be achieved before element-to-element contact can occur and result in wetting and dissolution.

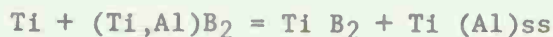
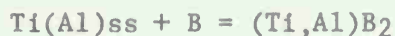
Reaction Bond

The reaction bond occurs when a new chemical compound is formed at the interface, such as the formulation of titanium diboride at the interface between boron and titanium.

⁽¹⁾React is restricted to those systems that result in the formation of a new chemical compound or compounds.

Exchange Reaction Bond

This is a special case of the reaction bond in which two or more reactions may occur: For example, the reaction between titanium-aluminum Ti(Al) solid solution and boron may be described as taking place in two steps:



Oxide Bond

The oxide bond may not involve new principles other than those enunciated earlier, but in the absence of detailed studies of bonding mechanisms, it appears desirable to introduce this bond type in a separate grouping. It may appear to be purely mechanical such as the silver-alumina whisker bonds studied by Sutton (1966) (ref 4). However, Moore (1969) (ref 4) showed that introduction of traces of oxygen converted the nickel-alumina bond to a reaction bond by formulation of the $\text{NiO Al}_2\text{O}_3$ spinel. Another example may be the bond formed between the oxide-coated surfaces of aluminum and boron by solution or reaction between the two oxides. The product exists as an oxide film at the interface and constitutes the bond in this pseudostable Class I composition system.

Mixed bond

This may be one of the most important categories. Breakdown from one type to another will be one source of mixed bonds such as the partial transition from a pseudo-Class I system to a Class II or Class III system.

Mechanical Aspects of the Interface

Nature of the Interface

To gain an understanding of the mechanical aspects of the interface in fiber composite materials, some attention must be given to the basic nature of the interface itself. The unique nature of the interfaces in fiber composite materials and the concomitant specific mechanical interactions produced at them, constitute one of the major factors in giving fiber composite materials their special properties.

The Nature and Effects of Residual Stresses at Composite Interfaces

The role of residual stresses is often ignored in analytical and experimental considerations of interfacial effects in composite materials. This oversight is unfortunate because the resultant interpretation of properties and behaviors is usually misleading. Residual stresses are an inherent characteristic of composite materials. The primary origin of residual stress in fiber composites is twofold: thermal and mechanical.

Thermal origin is the most prevalent, arising from the differing thermal coefficients of expansion of the component materials, since composites are invariably used at different temperatures than those at which they are fabricated. The differing thermal expansions or contractions of the fiber and matrix set up thermally induced stresses when cooled down from the compositing temperature. Metal matrix composites, particularly, are fabricated at temperatures which are quite high relative to the ambient, and thus hold the possibility of producing very high stress levels.

The second main source of composite residual stresses is the difference in flow stress⁽²⁾ between components. This is important when the composite is subjected to mechanical deformation at a level where one or more of the component materials begin to flow plastically. Under these conditions, the residual stresses developed upon loading the composite stem from the different amounts of plastic flow which have taken place among components of the composite.

MICROMECHANICS

Introduction

On structural composites, fibers are stiff and strong and serve as the load-bearing constituent. The matrix surrounding the fibers is soft and weak, and its direct load bearing capacity is negligible. However, the role of matrix is very important for the structural integrity of composites; matrix protects fibers from hostile environments and localizes the effect of broken fibers. Typical properties of some fibers are listed in Table 3 (ref 5).

In discussing composite properties, it's important to define the volume element which is small enough to show the microscopic structural details, yet large enough to represent the overall behavior of the composite. Such a volume element is called the representative volume element. A simple representative volume element can consist of fiber embedded in a matrix block.

Density of Composites

Consider a composite of mass M and volume V . Here V is the volume of a representative volume element. Since this composite is made of fibers and matrix, mass M is the sum of the total mass M_f of fibers and M_m of matrix.

$$M = M_f + M_m \quad (1)$$

The composite volume V includes the volume V_v of voids.

$$V = V_f + V_m + V_v \quad (2)$$

⁽²⁾Flow Stress: stress necessary to propagate plastic deformation, once initiated .

Dividing Eq (1) (2) by M and V, respectively, leads to the following relations for the mass fraction and volume fractions.

$$m_f + m_m = 1 \quad (3)$$

$$v_f + v_m + v_v = 1 \quad (4)$$

In this section, the subscript f, m, v are exclusively used to denote fiber, matrix and void respectively. The composite density ρ follows from (1) and (2) as:

$$\rho = \frac{M}{V} = \frac{\rho_f V_f + \rho_m V_m}{V} = \rho_f v_f + \rho_m v_m \quad (5)$$

In terms of mass fractions, ρ becomes:

$$\rho = \frac{1}{m_f/\rho_f + m_m/\rho_m + v_v/\rho} \quad (6)$$

Eq (6) is frequently used to determine the void fraction:

$$v_v = 1 - \rho \left(\frac{m_f}{\rho_f} + \frac{m_m}{\rho_m} \right) \quad (7)$$

Composite Stresses and Strains

In Eq (5) the composite density ρ is seen to be equal to the densities of the constituents averaged over the composite volume. The composite stresses and composite strains are defined similarly.

Suppose the stress field in the representative element is σ_i , the composite stress $\bar{\sigma}_i$ is defined by

$$\bar{\sigma}_i = \frac{1}{V} \int_V \sigma_i dV = \frac{1}{V} \left(\int_{V_f} \sigma_i dV + \int_{V_m} \sigma_i dV + \int_{V_v} \sigma_i dV \right) \quad (8)$$

$$\frac{1}{\rho} = \frac{V}{M} = \frac{V_f + V_m + V_v}{M} = \frac{M_f/\rho_f + M_m/\rho_m + V_v}{M}$$

Since $m_f = M_f/M$ and $m_m = M_m/M$ from (1) & (3)

$$1/\rho = m_f/\rho_f + m_m/\rho_m + V_v/M$$

$$\text{Since } M = \rho V, \quad V_v/M = V_v/\rho V = v_v/\rho$$

$$\text{Therefore: } 1/\rho = m_f/\rho_f + m_m/\rho_m + v_v/\rho$$

$$\text{And } \rho = 1/(m_f/\rho_f + m_m/\rho_m + v_v/\rho)$$

We now introduce the volume average stress $\bar{\sigma}_{fi}$ and $\bar{\sigma}_{mi}$ in the fibers and matrix, respectively,

$$\begin{aligned}\bar{\sigma}_{fi} &= \frac{1}{V_f} \int_{V_f} \sigma_i dV \\ \bar{\sigma}_{mi} &= \frac{1}{V_m} \int_{V_m} \sigma_i dV\end{aligned}\tag{9}$$

since no stress is transmitted in the voids (i.e. $\sigma_i = 0$ within V_v), Eq (8) can be written as

$$\bar{\sigma}_i = V_f \bar{\sigma}_{fi} + V_m \bar{\sigma}_{mi}\tag{10}$$

Similarly to the composite stress, the composite strain is defined as the volume-average strain, and is obtained as:

$$\bar{\epsilon}_i = V_f \bar{\epsilon}_{fi} + V_m \bar{\epsilon}_{mi} + V_v \bar{\epsilon}_{vi}\tag{11}$$

Unlike the stress, the strain in voids does not vanish. The void strain is defined in terms of the boundary displacements of the voids. However, since the void fraction is usually negligible i.e., less than 1%,

$$\bar{\epsilon}_i = V_f \bar{\epsilon}_{fi} + V_m \bar{\epsilon}_{mi}\tag{12}$$

with understanding that $(V_f + V_m)$ is unity.

Note that Eq (10) and (11) simply follow from the definitions of the composite stress and strain that the composite variables are the volume average. Thus these equations are valid regardless of the material behavior.

Rule of Mixture

Predictions of the response of a unidirectionally reinforced composite were based initially upon postulated states of stress and load transfer mechanism. The outgrowth of these analyses was the Rule-of-Mixture.

By assuming an isostrain criterion, i.e., both fiber and matrix are strained equally and uniformly, the longitudinal and transverse strength, stiffness, and the major and minor Poisson's ratio are found by the use of parallel and series spring models. The assumptions are:

- o Elastic and plastic isotropy
- o Perfect mechanics continuum at interface
- o No chemical reaction between constituents
- o An absence of residual stress
- o An absence of rheological interaction at interfaces.
- o Definability of in-situ properties of both fiber and matrix properties at the strains in question.

Experimental data have shown that only the longitudinal modulus and major Poisson's ratio can be reliably predicted by these simple approximations. Kelly and Davies (ref 2) reported these approximations as:

$$E_{11} = E_f v_f + E_m (1 - v_f) \quad (13)$$

$$\nu_{12} = \nu_f v_f + \nu_m (1 - v_f) \quad (14)$$

where E_{11} = Young's modulus of the composite parallel to the fibers

E_f = Young's modulus of the fiber

E_m = Young's modulus of the matrix

ν_{12} = major Poisson's ratio of the composite

ν_f = Poisson's ratio of the fiber

ν_m = Poisson's ratio of the matrix

v_f = Vol % of fiber in the composite

According to the rule-of-mixtures prediction, the longitudinal strength of the composite would be:

$$\sigma_{11} = \sigma_f v_f + \sigma_m (1 - v_f) \quad (15)$$

where σ_f = stress in the fiber

σ_m = stress in the matrix, at the fracture strain of the composite.

This prediction of longitudinal strength in Eq 15 is the one that has proven so useful in practice. The previous equations give properties such as the modulus which are rather insensitive to important factors such as the nature of the filament or matrix interface.

Strength

Unidirectional composites possess excellent strength and stiffness in the longitudinal direction because load is carried mostly by fibers. In the other loading conditions, the load sharing is about equal between fibers and matrix; therefore, composite strengths are comparable to those of the matrix used. Another parameter which plays a very important role in the strength of composites is the interface between fiber and matrix.

Failure of a material is initiated at the weakest point. A weak interface will certainly lead to a premature failure when a substantial load sharing is expected at the interface.

Load sharing by constituent phases depends on the type of loading. Therefore, we shall discuss strengths of unidirectional composites under five different loadings: longitudinal tension and compression, transverse tension

and compression, and shear.

Consider a unidirectional composite subjected to a unidirectional tension in the fiber direction. Since $\bar{\epsilon}_x = \bar{\epsilon}_{mx} = \bar{\epsilon}_{fx}$, Figure 2 (ref 5) schematically presents the stresses in the constituent phases. This figure has been constructed based on the following observations.

1st: Fiber is linear elastic up to fracture.

2nd: Matrix is linear initially, then nonlinear as strain increases.

The strain at which nonlinearity starts to appear is greater than the fracture strain of the fiber.

Since not all fibers are expected to be of equal strength and equally stressed, some fibers will fail before others. When these fibers break, there are three modes of further damage growth depending on the properties of the matrix and interface.

If the matrix is brittle and the interface strong, the cracks created by the fiber breaks will propagate through the matrix across the neighboring fibers, leading to the composite failure.

If the interface is weak, then interfacial failure can be initiated at the fiber breaks and the fiber matrix debonding will grow along the broken fibers. A longitudinal damage growth is also possible in the form of matrix yielding between fibers.

If the matrix is ductile with low yield stress, as far as the composite strength is concerned, the latter two modes of damage growth have a similar effect.

Therefore, we shall simply divide the failure mode under a longitudinal tension into the transverse crack propagation mode and the longitudinal damage growth mode. The transverse crack propagation mode is in fact what is observed in brittle, homogeneous materials. In this failure mode, the strength of stronger fibers cannot be fully utilized, and hence the composite strength is not optimum.

In the other extreme case of complete longitudinal damage growth mode, broken fibers are simply separated from intact ones as far as load sharing is concerned, and the composite behaves like a dry bundle of fibers.

CONSOLIDATION PROCESSES

A wide variety of metallurgical processes have been employed for the fabrication of filament reinforced metal matrix composites. Major attention has been placed on diffusion bonding and more recently, spray bonding as the preferred methods.

Each process employed to consolidate a composite structure requires specific methods to deal with the highly anisotropic materials and relatively fragile reinforcing filaments, thus differing from conventional metallurgical techniques. Each process employed must meet to some reasonable degree the following objectives:

- o Incorporate the filament without breakage.
- o Consolidate the composite with minimal filament degradation.
- o Establish and maintain filament alignment.
- o Achieve a high density matrix.
- o Offer variable filament volume loading.
- o Establish filament-matrix interfacial bond sufficient to transmit applied load from matrix to filament.
- o Allow for post fabrication heat treatment.
- o Allow a degree of matrix selection and alloying.
- o Give flexibility in filament spacing.
- o Some methods must provide capability for cross and angle ply lay-ups.
- o Minimize product variability.
- o Be amenable to scaling up.

The consolidation processes used for metal matrix composites are: diffusion bonding, plasma spray bonding, electro forming, liquid metal infiltration, high energy rate forming and hot rolling bonding.

Diffusion Bonding

The most widely used method of consolidation of metal matrix composites is simultaneous application of heat and pressure, known as diffusion bonding. It is a static pressure process as differentiated from high energy methods which use dynamic pressure techniques. This method is used commonly to consolidate B/Al, Borsic/Al, Ti/SiC and Ti/Borsic composites.

The diffusion bonding process for a single uniaxial filament reinforced metal matrix is shown schematically in Figure 3 (ref 6).

Using diffusion bonding as the most advanced consolidation process, marked improvements in the properties of B/Al composites have been achieved. Optimizing the fabrication parameters, utilizing the wide diameter 5.6 mil boron filament, and heat treating the composite material have produced excellent results. Maximum values for 48 V/O 5.6 mil B/6061 Al (uniaxial reinforced, heat treated) have been reported as 220,000 psi longitudinal tensile

strength, over 40,000 psi transverse tensile strength, and 625,000 psi compression strength.

Plasma Spray Bonding

Plasma spray bonding to prepare monolayer tapes for consolidation by diffusion bonding into thicker sections, has become an important consolidation process. The rate of powder feed affects the spray deposit in terms of deposit efficiency and quality of matrix filament bond achieved. Good results were achieved at feed rates of 3 lb/hr of metal powder, 240 to 400 mesh, at 4 to 5 inches torch to substrate.

Electroforming

Electrodeposition in the fabrication of metal matrix composites was one of the first methods studied because the composite could be formed at low temperature, thus minimizing the degradation of reactive filaments in metal matrices. A schematic of the electroforming process is shown in Figure 4 (ref 7).

Two practical limitations which seem to be inherent in the electroforming process are:

- o The impurities which are incorporated into the composite from the bath during deposition.
- o The inability to deposit alloys from solution.

A number of advantages can be cited for this process, including:

- o A low room temperature fabrication process.
- o A high density monofilament matrix can be obtained.
- o Excellent filament matrix interface contact is possible.
- o Filament spacing is accurately controlled.
- o Volume loading of reinforcement is quite flexible.
- o A variety of mandrell shapes can be accommodated.

Due to its inherent limitations, this method has marginal utility and serious problems for most metal matrix composites.

Liquid Metal Infiltration

Several model refractory metal systems have been consolidated in this manner such as tungsten reinforced copper, where little or no mutual solubility exists. The tungsten wires collimated in a ceramic tube were infiltrated by liquid copper. The technique is limited and had not received wide use because

of the few reinforcements which are stable in molten metals. This technique has been employed on Al_2O_3 whisker reinforced composites. The problem with the Al_2O_3 whisker in matrices such as aluminum and nickel has not been whisker degradation reactions, but rather the proper wetting of the whisker with the matrix. The wetting is controlled by various coating, matrix alloying, and control of the infiltration atmosphere. This process is shown in Figure 5 (ref 7).

High Energy Rate Forming

Very high pressure pulses of short duration have been utilized to fabricate composites in a method known as high energy rate forming. The method draws its advantage from the instantaneous application of very high pressures for very short time periods which prevents interaction of matrix and filament or whisker reinforcement while providing densitification of reasonably complex shapes. The typical conditions for this process are applied pressures of 400,000 psi and energy pulse duration of micro-to-milliseconds. A schematic process is shown in Figure 6 (ref 7).

Hot Rolling Bonding

The simultaneous application of heat and pressure in the diffusion bonding process has been modified to provide for a short reaction time during fabrication in a process known as hot rolling bonding. This method has been extensively employed by Metcalfe and his co-workers to counter the reacting problem with a reactive system such as BTi. The roll bonding method provides for a very short time of contact between the filaments and matrix at bonding temperature, but is practically limited to tapes as monolayers or a few layers thick.

A major problem in hot rolling bonding is the difficulty in obtaining high volume loading of the filaments. Another is cross- and angle-ply configurations which, for diffusion bonded sheets in alternate directions, are reasonably straight forward. For thin tapes this is a big problem in consolidation. The schematic process is shown in Figure 7 (ref 8).

MECHANICAL PROPERTIES OF COMPOSITE MATERIALS

Strength

The strength of unidirectional and multidirectional composites can be determined by quadratic interaction failure criteria in stress and strain space.

Failure Criteria

For the determination of strength of any material, it is the usual practice to estimate the stress at the time and location when failure occurs. In the case of conventional materials, we need only to determine the maximum tensile, compressive or shear stress and can make some observation about the failure and failure mechanism. This process is relatively straightforward because isotropic materials have no preferential orientation and usually one strength constant will suffice. The isotropic material is essentially a one dimensional or one

constant material. The Young's modulus for stiffness will suffice because Poisson's ratio is taken to be about 0.3 and the uniaxial tensile strength will also suffice because the shear strength is taken to be about 50 to 60% of the tensile.

For the composite materials, however, the one constant approach for stiffness or for strength is no longer adequate. We saw earlier that four elastic constants were needed for the strength of unidirectional composites. We know that unidirectional composites have highly directionally dependent strengths. The longitudinal strength can be twenty times that of the transverse and shear strength. So we cannot say quickly that specific stress components are responsible for the failure.

The determination of strength using failure criteria is based on the assumption that the material is homogeneous (properties do not vary from point to point) and its strength can be experimentally measured with simple tests. Failure criteria provide the analytic relation for the strength under combined stresses. There is another approach for strength determination, using fracture mechanics. A material is assumed to contain flaws. The dominant flaw based on its size, shape, and location determines the strength when its growth cannot be stopped.

For composite materials, we need a failure criterion for the unidirectional plies. The strength of a laminated composite will be based on the strength of the individual plies within a laminate. We would expect successive ply failures as the applied load to a laminate increased. We will have the first ply failure (FPF) to be followed by other ply failures until the last ply failure which would be the ultimate failure of the laminate. The ply stress and ply strain calculations for symmetric and general (random) laminates are intended for strength determination.

There are two popular approaches for failure criteria of unidirectional composites. They are all based on the on-axis stress or strain as the basic variable.

Maximum Stress and Strain Criteria

$$\sigma_x \leq X, \quad \sigma_y \leq Y, \quad \sigma_z \leq S \quad (16)$$

Failure occurs when one of the equalities is met. Using the linear relation we can express the equation above in the maximum strain criterion:

$$\epsilon_x \leq X/E_x, \quad \epsilon_y \leq Y/E_y, \quad \epsilon_z \leq S/E_z \quad (17)$$

Failure occurs when one of the equalities is met. These two criteria are not the same. Only when Poisson's ratio of the unidirectional material is zero, does the criteria become identical; conceptually they are similar. Each component of stress or strain has its own criterion and is not affected by the

components. There is no interaction.

Quadratic Interaction Criterion

$$F_{ij} \sigma_i \sigma_j + F_i \sigma_i = 1 \quad (18)$$

This can be expressed in strain components:

$$G_{ij} \epsilon_i \epsilon_j + G_i \epsilon_i = 1 \quad (19)$$

Failure occurs when either equation is met.

Failure criteria serve important functions in the design and sizing of composite laminates. The criteria are not intended to explain the mechanisms of failure. Failure in composite materials involve many modes: viz., fiber failure, matrix failure, delamination, interfacial failures and buckling. Furthermore, the various modes interact and can occur concurrently and sequentially.

Quadratic Failure Criterion

Eq 18 can be expanded for the case of two-dimensional stress or $i, j = 1, 2, 6$.

$$F_{11} \sigma_1^2 + 2F_{12} \sigma_1 \sigma_2 + F_{22} \sigma_2^2 + F_{66} \sigma_6^2 + 2F_{16} \sigma_1 \sigma_6 + 2F_{26} \sigma_2 \sigma_6 + F_1 \sigma_1 + F_2 \sigma_2 + F_6 \sigma_6 = 1 \quad (20)$$

since unidirectional composite is in its orthotropic axis, as shown in Figure 8 (ref 5).

The strength should be unaffected by the direction or sign of the shear stress component. If the shear stress is reversed, the strength should remain the same.

Sign reversal for the normal stress components, say from tensile to compressive, is expected to have a significant effect on the strength of composite. Thus, all terms in Eq 20 that contain linear or first-degree shear strength must be deleted from the equation. There are three such terms:

$$F_{x5} \sigma_x \sigma_5, F_{y5} \sigma_y \sigma_5, F_5 \sigma_5 \quad (21)$$

Since the stress components are in general not zero, the only way to ensure that the terms above vanish is for

$$F_{x5} = F_{y5} = F_5 = 0 \quad (22)$$

With the removal of the three terms, Eq 20 can be simplified:

$$F_{xx} \sigma_x^2 + 2F_{xy} \sigma_x \sigma_y + F_{yy} \sigma_y^2 + F_{55} \sigma_5^2 + F_x \sigma_x + F_y \sigma_y = 1 \quad (23)$$

Of the six material constants or strength parameters, five can be measured by performing simple tests.

Longitudinal Tensile and Compressive Tests

Let X = longitudinal tensile strength

X' = longitudinal compressive strength

These strengths are measured by uniaxial tests shown in Figure 9 (ref 4).
Substituting the measured strength into Eq 23.

$$\text{If } \sigma_x = X \quad (24)$$

$$F_{xx} X^2 + F_x X = 1$$

$$\text{If } \sigma_x = -X' \quad (25)$$

$$F_{xx} X^2 - F_x X' = 1$$

So we can get

$$F_{xx} = 1/X X', \quad F_x = 1/X - 1/X' \quad (26)$$

Transverse Tensile and Compressive Tests

Let Y = transverse tensile strength

Y' = transverse compressive strength

$$F_{yy} = 1/Y Y', \quad F_y = 1/Y - 1/Y' \quad (27)$$

Longitudinal Shear Stress

Let S = longitudinal shear stress

$$F_{ss} = 1/S^2 \quad (28)$$

So we obtained five of six coefficients in failure criterion of Eq 23. The one remaining term is related to the interaction between the two normal stress components. The only way that these coefficients can be measured is both normal stress components to be non zero; this requires a combined stress biaxial test. This experimental task unfortunately is not easy to perform as the simple uniaxial or shear test.

Strength data for other unidirectional composites is in Table 4 (ref 5) and the strength parameters are listed in Table 5 (ref 5) and Table 6 (ref 5) for stress and strain spaced representations of the strength parameters, respectively.

Tensile Properties

Stress and Strain Behavior

The tensile properties of state of the art boron-reinforced aluminum and titanium are summarized in Table 7 (ref 2). Typical stress-strain diagrams for unidirectionally reinforced B/Al composites tested parallel (0°) and perpendicular (90°) to the direction of reinforcement are shown in Figure 10 (ref 9).

There are several interesting features of these stress-strain curves which deserve special mention. If the 0° curve is critically examined three distinct regions are noted, as schematically shown in Figure 11 (ref 10). With the initial application of load, both phases deform elastically with the rule-of-mixture relation giving a fairly accurate prediction of modulus. Eventually, the yield strength of the matrix is exceeded and the matrix begins to flow plastically. As a result, the matrix contribution to composite stiffness is substantially reduced. Composite stiffness at a given strain is then determined by the weighted average of the modulus of the reinforcement and instantaneous strain hardening rate of the matrix, $d\sigma/d\epsilon$. In the case of B/Al, $d\sigma/d\epsilon$ for aluminum is negligibly small compared with the 59×10^6 psi modulus of boron filament. The slope of the stress-strain curve in this region is referred to as the system secondary modulus and is generally 70 to 90% of the initial modulus. The stress-strain behavior in this second region is, of course, neither elastic nor linear.

The system experiences permanent deformation by reason of matrix flow and breakage of severely weakened filaments, and the strain hardening rate of matrix is not necessarily constant, although changes in matrix hardening rate have little influence on the composite stress-strain behavior in this region. This second stage continues until filament breakage is encountered, whereupon the slope of the stress-strain diagram is again observed to decrease (stage III), eventually resulting in composite fracture.

If the yield strain of the matrix exceeds the fracture strain of the brittle filament, stage II - type behavior will not be observed.

The transition from stage I and stage II depends on the yield strength (or strain) of the matrix and the magnitude of residual consolidation stresses. In the case of ductile metal wire reinforcement, Stage III is extended because both phases are capable of deforming plastically, causing failure to be postponed.

A rule-of-mixture type prediction of composite tensile strength will require knowledge of the stress-strain behavior of each component under the conditions it would experience in the composite. This includes such subtle points as matrix grain size, impurity distribution, consolidation induced defects, the interdiffusion of components, and the presence of reaction products. Filament strength and strain-to-failure must reflect the chemical and/or mechanical degradation resulting from consolidation and forming operations.

Since filament tensile strength is sensitive to gage length, it is very conservative to use filament strengths measured on 1 inch gage lengths in composites where the critical load transfer length is generally less than 0.1 inch. The influence of closely spaced, stiff and strong filaments on the flow stress of the matrix and the possibility of chemical change as a result of reaction with the filaments must be considered.

Residual consolidation stresses will influence the response of both phases to applied loads. Using the isostrain criterion, the contribution of each phase to the composite 0° tensile strength will depend on the strength developed in the two phases at the flow fracture strain of the brittle filament. Composite failure strains are generally between 0.3 and 0.6%. Anything which reduces the strain-to-failure of the filaments will be directly evidenced in reduced composite 0° tensile strength.

Residual Stress

Residual consolidation stresses arise because of the thermal expansion mismatch of two components. For instance, the thermal expansion coefficient of boron is 2.8 micro-in/in/ $^\circ\text{F}$ while 6061 aluminum is 13.1 micro-in/in/ $^\circ\text{F}$ and titanium 6 Al-4V is 4.7 micro-in/in/ $^\circ\text{F}$.

Since consolidation of filament and metal takes place at relatively high temperatures (1000 $^\circ\text{F}$ for B/Al and 1500 $^\circ\text{F}$ for B/Ti), these differences in expansion coefficient will result in the formation of longitudinal and radial residual compressive stresses on the filaments and corresponding tensile stresses in the matrix. Quantitative estimates of the magnitude of these residual stresses have been made, but these estimates are complicated by the relaxation of the matrix both during cooling from the consolidation temperature and at room temperature after consolidation.

In most cases one finds that matrix yielding takes place during cooling so that in the absence of relaxation the matrix can exhibit no elastic deformation when the composite is loaded in tension. Since elastic behavior is observed one must conclude that matrix relaxation has taken place. If the filament-matrix bond strength is inadequate, cooling will result in slippage along this boundary instead of causing matrix strain hardening, and the magnitude of the resultant residual stresses will be reduced. This condition, however, has never been observed in systems of practical interest.

Low composite failure strains have been observed for these cross-and angle-ply materials using normal consolidation process. A technique which has helped alleviate this matrix triaxial tensile stress state is a low temperature quench. Cooling the composites below room temperature cause further matrix flow to help accommodate the thermal expansion mismatch. Subsequent heating to room temperature causes a relaxation of the residual stresses and an increase in composite failure strain.

Transverse Tensile Properties

There is a striking difference between the 0° and 90° tensile properties of these unidirectionally reinforced materials (Table 7 and Figure 9). The tensile modulus, strength, and ductility are all lower for transverse tensile tests than longitudinal tensile tests. In fact, there is a 10:1 difference in strength and a 1.5:1 difference in modulus. The lower moduli and strength are, in part, due to the fact that the isostrain criterion no longer applies. That is, the matrix is free to flow nearly independently of the filaments. Under these conditions it becomes more difficult to predict composite stiffness and strength. If, however, the filaments are well bonded to the matrix and free of defects, the transverse strength should approach or exceed the strength of the bulk matrix alloy. The slight strength reduction can be related to structural imperfections in composites and the splitting of very weak boron filaments.

Quantitative examination of the fracture surfaces indicates that in all cases the relative area of split filaments on the fracture is less than the Vol % boron in the composite.

Failure is controlled by matrix properties. This contrasts with composites where the matrix has been heat treated and aged to a T-6 condition. The matrix is now substantially stronger than the majority of the boron filaments (i.e. 45,000 psi for T-6 6061 Al compared with 30,000 psi mode for boron filament splitting).

Here the transverse strength is found to be a fairly sensitive function of boron content as would be expected if filament splitting controls the composite failure. Generally, the relative area of split filaments on the fracture surface is found to be larger than the Vol % boron, which clearly indicates that the filaments offered the least resistance to crack propagation and probably control composite failure. The higher transverse strengths of the heat treated system at all reinforcement levels reflect the high strength and stress concentration accommodation ability of the matrix.

Improvement of the transverse properties of unidirectionally reinforced B/Al can be accompanied by development of improved filaments, by third phase additions, or by heat treatment of the matrix. An alternate solution is to use cross-or angle-ply B/Al where substantial off-axis loads are anticipated. The use of cross-or angle-ply:

- o imposes a reduced upper limit on the volume fraction of boron filament since interpenetration of adjacent boron layers becomes impossible,
- o increases the probability of introducing filament defects during consolidation, and
- o introduces complex shear forces between adjacent boron layers.

None of these features in themselves would eliminate cross-or angle-applied composites from active consideration but they do introduce complexities which, in most cases, could be avoided if the transverse strength of unidirectional composites were increased to 40 - 50,000 psi. The first of the solutions, improved filaments, is being pursued by the development of the larger diameter (8 mil) boron filament, by the use of different boron filament substrates (carbon), and by the development of additional varieties of filament (SiC, Al₂O₃, and graphite). The filament development is usually, however, a long and costly process. The second solution, third phase additions, shows great promise. Composites have been fabricated and tested with minor (5-15 Vol%) 90° stainless steel wire additions and with layers of titanium foil which demonstrate substantial improvements in transverse properties without materially affecting longitudinal properties. Matrix heat treatment, in addition to modifying the matrix strength and its susceptibility to stress concentrations, can also be used to change the residual stress state in the composite. Because of the higher thermal expansion coefficient of aluminum and titanium relative to boron, rapid cooling will result in residual radial compressive stresses on the boron filament which may increase composite transverse strength and ductility in those systems currently limited by filament splitting.

Compression

The ultimate strength of composites tested in compression has been found to equal or exceed their ultimate tensile strength. The modulus of elasticity is nearly identical in tension and compression. Typical stress - strain diagrams for 0° and 90° specimens tested in compression are shown in Figure 12 (ref 11).

The compression of 0.20 x 0.25 x 0.75 inch specimens of unidirectionally-reinforced 50 Vol % Borsic /Al parallel to the filaments (Figure 12) resulted in a modulus of 34×10^6 psi and a compressive strength of 297,000 psi with failure occurring as a result of the "brooming" of one end of the specimen.

In comparison, a similar sized specimen tested (Figure 13) (ref 12) at 90° to the filaments resulted in a modulus of 20×10^6 psi and an ultimate compressive strength of 37,000 psi. The 0° compressive strength is very much more sensitive to boron content than the 90° strength, the transverse compressive strength being determined primarily by the shear strength of the matrix.

Impact

The impact energy for 50 Vol % Borsic - aluminum composites, as determined from full size Charpy "V" notch specimens is shown in Figure 14 (ref 13). The LT notch-filament configuration not only yields the highest impact energy but also demonstrates an increase in energy absorption capacity with increasing boron content. In comparison, the TT and TL notch-filament configurations have much lower energy absorption capacity and appear to be insensitive to boron content. In the case of LT notch filament configuration, the crack front is propagating normal to the filaments with composite failure requiring the fracture of all filaments in the cross section. With this configuration, cracks can be deflected parallel to the filaments along the filament-matrix interface, thereby increasing the composites energy absorption capacity. The stresses normal to

the filament are considerably smaller than is the case for the other two configurations. In contrast, cracks propagating in the TT and TL specimens are not required to fracture all the boron filaments in the cross section. Instead, the high transverse stresses acting on the filament near the crack tip cause the filament to split longitudinally. The stress required to cause longitudinal filament splitting (30,000 psi) is considerably less than the stress required for the fracture of boron filaments normal to their axis (450,000 psi).

Although the strengths cited above cannot be directly applied to impact energy predictions, they do indicate that the filament will be able to make a larger contribution to the composites energy absorption capacity in the LT configuration as compared with the TT and TL configuration. In addition, longitudinal filament splitting can provide an easy path for crack propagation in the case of the TT and TL configuration. The slight improvement of the TT configuration relative to the TL configuration may result from a larger matrix contribution to impact energy in the TT case. More extensive plastic deformation is possible around the filaments which run parallel to the crack front in the TT specimen. In comparison, relatively little plastic deformation can take place in the TL specimen without transferring load into the filaments, since the slip bonds will necessarily intersect the filament.

Elevated Temperature Tensile Strength

The variation of ultimate tensile strength of unidirectional B/Al composites with test temperatures is shown in Figure 15 (ref 12). For comparative purpose, the variation of tensile strength with temperature for 6061 aluminum is also shown. The most important observation is that the composites retain their strength exceptionally well up to about 600°F. At 600°F the composite tensile strengths are still 10 to 30 times higher than the tensile strength of the aluminum alloy matrix. The variation of ultimate tensile strength of cross and angle ply B/Al composites is shown in Figure 16 (ref 12). Notice that the +5° angle-ply and 0° to 90° cross-ply composites have the same temperature dependence as the unidirectional composites shown in Figure 16, while the +30° angle-ply composites reflect the decreasing matrix strength above 300°F.

The temperature dependence of boron filament strength is not presented in Figure 15, but strength reductions of 20 to 40% have been reported from room temperature tensile strength of approximately 500,000 psi. The observed composite strengths at 750°F are less than rule-of-mixtures predictions, even if a 40% filament strength reduction is assumed.

The direct application of filament tensile data to composite strength predictions is often times quite misleading, particularly at elevated temperatures where the chemical reactivity of the filament with the atmosphere or metal matrix introduces an additional complexity.

The variation of transverse tensile strength with test temperature for 25, 37 and 50% B/6061 Al composite is shown in Figure 17 (ref 12). The temperature dependence of the transverse strength of these composites is independent of the reinforcement content.

IMPROVED MECHANICAL PROPERTIES

Introduction

Wider diameter filaments not only have increased transverse strength, but also improved reliability of their longitudinal strength. For example, the amount of 5.6 mil boron filaments below 400,000 psi longitudinal tensile strength is only 10% compared to 20% of the 4.0 mil boron filament. This is perhaps more significant than the increase in average tensile strength by about 25,000 psi since it is the weak filaments that will fail first, and perhaps initiate premature composite failure. Filament splitting has also been substantially reduced, eliminating a major problem in premature failure.

Improved Filament

To avoid adverse reactions between the boron filament and metals during high temperature fabrication of metal matrix composites, the surface of the boron filament is coated with a diffusion barrier of boron carbide (B_4C). Excellent filament properties are maintained in both aluminum and titanium matrix composites. The strength of the B_4C coated boron filament is superior to the boron and the SiC coated boron filament. A typical set of filament test data shows that the strength of boron is increased by the addition of a layer of B_4C (Figure 18) (ref 6). The average strength for boron is over 500 ksi, increased to over 600 ksi for B_4C coated boron and reduced to below 500 ksi for silicon carbide boron.

A low cost silicon carbide filament is developing for the reinforcement of metal matrix composites. The silicon carbide filament is potentially very low cost and is compatible with high temperature processing in aluminum and titanium. Its tensile strength is 500 ksi, modulus is 62 Msi and density is 0.11 lb/in³.

Improved Composites

The properties of B/Al composites as reported by various commercial sources have been compiled in Table 8 (ref 2). There are two trends that should be noted. One is that the wide diameter filaments give consistently higher composite strengths in both the longitudinal and transverse orientations. The other is that heat treatment to the T-6 condition for an aluminum alloy with wide diameter filament improved the properties significantly. The maximum values of 230,000 psi for longitudinal tensile strength of B/6061-T6 Al and 45,000 psi for transverse tensile strength of B/2024-T6 Al are indeed encouraging.

The properties of several B/Ti composites are given in Table 9 (ref 2). Here, the excellent longitudinal and transverse strengths of the composite with large diameter boron are easily seen. The value of 185,000 psi and 64,000 psi respectively, for a B/Ti-6Al-4V composite are quite promising. The transverse tensile strength of the titanium composite containing wide diameter boron is markedly superior, just as was the result for aluminum matrix composites.

A critical property evaluation of the best available aluminum and titanium alloy matrix composites with boron filament reinforcement has recently been completed by Kreider, Dardi and Prewo (ref 14). This investigation has produced an important body of information on composite behavior including bonding conditions, environmental effects, transverse properties, off-axis properties, failure mechanisms in fatigue, notch bending fracture, and notched tensile fracture behavior. A few of their major findings will be considered here. Longitudinal tensile strengths for 5.6 mil B with various Al alloys are shown in Table 10 (ref 14). A very high volume loading was achieved, as high as 70 Vol% B in 2024 Al. Thus, strengths to nearly 280,000 psi were obtained with an elastic modulus value of 40×10^6 psi.

Longitudinal strengths for Borsic/Al are listed in Table 11 (ref 14). Here, volume loadings of about 60% Borsic have been obtained with strengths exceeding 200,000 psi and moduli approaching 40×10^6 psi.

The transverse tensile strengths for 5.6 mil B in various Al alloys are shown in Table 12 (ref 14). The transverse strength of B/2024-T6 Al in this case is almost 50,000 psi for a 45 Vol % loading. The heat treatment to the T6 condition again much improves the values. This importance of T6 treatment is again seen in Table 13 (ref 14) where an extensive series of tests is reported for 5.7 mil Borsic/Al composites.

The change in longitudinal tensile strength with temperature for a 5.6 mil B/6061 Al composite is given in Table 14 (ref 14). The decrease in strength with temperature is similar to that found for 4.0 mil boron composites. The transverse strength curve with temperature is much higher for the wide diameter filament reinforcement at low temperature in a heat treated matrix than for the smaller diameter filament or an untreated matrix.

APPLICATIONS

Introduction

The introduction of advanced composites began in 1965 with boron fiber and since that time rapid advances have been made toward application of these composites to military aircraft. The significant weight advantages and the success of these developments have led to production commitments on fighter empennage structures.

The potential for significant weight reductions in aerospace structures through the use of advanced composites was first realized on a broad scale by the military in the Air Force Project "Forecast" conducted in 1963. This observation was based on the then recent development of the high modulus, high strength, low density boron fiber and the superior mechanical properties that could be developed when these fibers were converted into composite laminates. Since that time, other filament have been developed that offer equal or increased potential for reduced structural weight, increased stiffness and lower cost.

The first pioneering work in fiber glass structure was carried out at the Wright-Patterson Air Force Base, Ohio, in the early 1940's. By 1943 a BT-15 fuselage of sandwich construction with fiber glass faces and balsa wood core had been static tested. Flight tests of this structure were performed March 1944. On a strength to weight basis, the fiber glass fuselage was 50% stronger than an aluminum structure. In May 1945 the first fiber glass reinforced AT-6C wing was fabricated.

Examples of Applications

The current state-of-the-art of advanced composite structures has entered the stage where limited application is beginning to be seen on some current production and prototype military aircraft systems, such as wing and fuselage section components, empennage, helicopter rotors, etc. Other examples are gear wheels for lubricating and coolant pumps, gear case housing for Allison T-56-A-18, golf club shafts, rackets, and fishing rods.

In the case of gear wheels, carbon reinforced plastics may be promising for such light weight, self lubricating gears. The potential advantages of fiber reinforced plastic gear wheels are as follows:

- o self lubricating, simplifying design,
- o will not rust and are chemically resistant,
- o reduced maintenance requirement,
- o low weight and, therefore, inertia,
- o high strengths and stiffness,

- o low coefficient of thermal expansion,
- o low coefficient of friction and low heat generated, and
- o low wear rate.

For the gear case housing, the Allison T-56-A-18 turboprop reduction gear assembly gear was fabricated by Goodyear Aerospace Corporation under a US Navy contract. Normally, the conventional material of construction is cast magnesium. The composite design was fabricated from 1 inch chopped boron-fiberglass reinforced epoxy, and provided the following advantages:

- o molded composite gear case as strong as the magnesium counterpart,
- o 13% lighter and approximately twice as stiff, and
- o eliminating internal ribs and adding external stiffness resulted in a reduction of parasitic power loss.

For such applications as above, elimination of costly machining and material wastage, precise dimensional control, and tailoring of strength and stiffness are further advantages of a molded advanced composite glass system.

Cost

In the case of B/Al, Alexander showed the cost history in Figure 19 (ref 15). For comparison, the cost analysis by Toth is given in Figure 20 (ref 16).

For boron the picture is further improved with the acceptance of a larger diameter filament. The 1971 average price for 4 mil boron filament was about \$210 per lb. With 5.6 mil boron a cost reduction of almost 50% is obtained. Further reductions with increasing volume are expected.

Given their high property levels, the advanced composite materials with wide diameter boron can now provide real competition to some aerospace alloys. Assuming a reasonable market exists, it is possible to project costs of \$50 per lb or less. These projections involve considerable guess work, but at least the general trend for B and B/Al has been steadily downward in price.

Further Applications

Present and future engineering design requirements demand a maximum of materials strength and stiffness at a minimum weight and potential for future use at temperatures above 1600° to 1800°F, as evidenced by the competition from the more familiar directions of metallurgical development of polymer matrix composites. In comparison to other composite matrices, the most obvious potential advantage of the metal matrix is its resistance to severe environments, high specific stiffness and strength, low thermal expansion, retention of strength at high temperatures, and high thermal and electrical conductivities.

In a composite structure, it is possible to emphasize environmental stability of the matrix at elevated temperatures, since the required mechanical strength and stiffness can be obtained from the reinforcement.

Currently, metal matrix composites are recognized for their tremendous design advantages to space system designers and users. However, as the property to cost ratio of these materials and the awareness and knowledge of their advantages continue to increase, metal matrix composites will be used in many other high performance land-based and aerospace applications.

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Table 1.^a Filament properties

Material	Tensile Strength (ksi)	Elastic Modulus (X 10 ⁶ psi)	Density (g/cm ³)	Density (lb/in. ³)	Strength/Density Ratio (X 10 ⁶ in.)
Beryllium	180	45	1.84	0.066	2.73
Molybdenum	320	48	10.2	0.369	0.88
Steel	600	29	7.9	0.286	2.10
Tungsten	575	59	19.3	0.697	0.83
E-Glass	500	10.5	2.5	0.092	5.43
S-Glass	600	12.5	2.5	0.092	6.52
Silica	500	10.5	2.2	0.079	5.43
Boron (on Tungsten)	450	58	2.6	0.096	4.68
Graphite ^b	285	50	1.66	0.060	4.75
Silicon Carbide (on Tungsten)	350	55	3.4	0.123	2.84
Alumina	350	70	3.98	0.144	2.43
Alumina Whiskers	2000	70	3.98	0.144	13.9
Silicon Carbide Whiskers	1500	70	3.2	0.115	13.0

^a There are many types varying in strength (150–375 ksi), modulus (25–75 X 10⁶ psi), and density (1.5–2.0 g/cm³).

Table 2. Boron type filament properties

Property	Unit	Boric		Boron	
		5.7 mil	4.2 mil	5.6 mil	4.0 mil
Diameter	10 ⁻³ in.	5.7 ± 0.1	4.0 ± 0.2	5.6 ± 0.1	3.9 ± 0.2
Ultimate tensile strength	10 ³ psi	450 – 475	425 – 450	450 – 500	450 – 500
Minimum tensile strength	10 ³ psi	300	300	350	335
Modulus of elasticity	10 ⁶ psi	59 – 60	57 – 60	59 – 60	57 – 60
Density	lb/in. ³	0.092	0.096	0.090	0.095
Length per pound	ft	34,500	57,000	38,000	70,000

Table 3.^b Typical fiber properties

Fiber	Diameter μ m	Density g/cm ³	Modulus GPa	Strength GPa
Graphite (T300, AS)	7	1.75	230	2.80
Boron (4-mil)	100	2.6	410	3.45
Glass (E)	16	2.6	72	3.45
Kevlar (49)	12	1.44	120	3.62

^aExtracted from C. T. Lynch and J. P. Kershaw, Metal Matrix Composites, Chemical Rubber Company (CRC), Boca Raton, FL, c1972. Permission granted by CRC.

^bExtracted from S. W. Tsai and H. T. Hahn, Introduction to Composite Materials, Technomic Publishing Company, Lancaster, PA, c1980. Permission granted by Technomic Publishing Company.

Table 4.^a Typical strengths of unidirectional composites in MPa

Type	Material	ν_f	Long. tens. X	Long. comp. X'	Trans. tens. Y	Trans. comp. Y'	Shear S
T300/5208	Graphite /Epoxy	0.70	1500	1500	40	246	68
B(4)/5505	Boron /Epoxy	0.50	1260	2500	61	202	67
AS/3501	Graphite /Epoxy	0.66	1447	1447	51.7	206	93
Scotchply 1002	Glass /Epoxy	0.45	1062	610	31	118	72
Kevlar 49 /Epoxy	Aramid /Epoxy	0.60	1400	235	12	53	34
	Aluminum		400	400	400	400	230

Table 5.^a Strength parameters in stress space for unidirectional composites

Type	Material	F_{xx} (GPa) ⁻²	F_{yy} (GPa) ⁻²	F_{xy} (GPa) ⁻²	$F_{\theta\theta}$ (GPa) ⁻²	F_x (GPa) ⁻¹	F_y (GPa) ⁻¹
T300/5208	Graphite /Epoxy	.444	101.6	-3.36	216.2	0	20.93
B(4)/5505	Boron /Epoxy	.317	81.15	-2.53	222.7	.393	11.44
AS/3501	Graphite /Epoxy	.476	93.48	-3.33	115.4	0	14.50
Scotchply 1002	Glass /Epoxy	1.543	273.3	-10.27	192.9	-.697	23.78
Kevlar 49 /Epoxy	Aramid /Epoxy	3.039	1572	-34.56	865.0	-3.541	64.46
	Aluminum	6.25	6.25	-3.125	18.90	0	0

Table 6.^a Strength parameters in strain space for unidirectional composites (dimensionless)

Type	Material	G_{xx}	G_{yy}	G_{xy}	$G_{\theta\theta}$	G_x	G_y
T300/5208	Graphite /Epoxy	12004	10680	-3069	11117	60.64	216.5
B(4)/5505	Boron /Epoxy	10374	27646	-2988	6961	129.6	214.3
AS/3501	Graphite /Epoxy	7363	7440	-1743	5821	39.22	130.76
Scotchply 1002	Glass /Epoxy	1913	18881	1712	3306	24.56	198.05
Kevlar 49 /Epoxy	Aramid /Epoxy	13453	47656	2068	4576	-149.8	350.8
	Aluminum	28387	28387	1976	13313	0	0

^aExtracted from S. W. Tsai and H. T. Hahn, Introduction to Composite Materials, Technomic Publishing Company, Lancaster, PA, 1980. Permission granted by Technomic Publishing Company.

Table 7.^a Tensile properties of boron/aluminum and boron/titanium composites

		50 v/o Boron-Aluminum		50 v/o Boron-Titanium	
		Base(psi)	Weight Normalized(in.)	Base(psi)	Weight Normalized(in.)
Ultimate Tensile Strength	0°	165,000	1.78 X 10 ⁶	155,000	1.22 X 10 ⁶
	90°	15,000	0.16 X 10 ⁶	30,000	0.24 X 10 ⁶
Tensile Modulus	0°	33.4 X 10 ⁶	356 X 10 ⁶	38.0 X 10 ⁶	300 X 10 ⁶
	90°	21.0 X 10 ⁶	224 X 10 ⁶	28.0 X 10 ⁶	220 X 10 ⁶

Table 8.^a Properties of 50 v/o boron/aluminum

Source	Filament	Matrix & heat treatment condition	Longitudinal tensile strength (ksi)	Transverse (ksi) tensile strength
A	Borsic/4.2 mil	6061-F; 2024-F	165	13
A	Borsic/4.2 mil	6061-T6; 2024-T6	180	20
B	Boron/4.0 mil	6061-F	180	13.5
B	Boron/4.0 mil	6061-T6	205	20
C	Boron/4.0 mil	6061-F	175	10
C	Boron/4.0 mil	6061-T6	170	15
A	Borsic/5.7 mil	6061-F	190	19
A	Borsic/5.7 mil	2024-F	190	27
A	Borsic/5.7 mil	6061-T6	200	36
A	Borsic/5.7 mil	2024-T6	200	45
B	Boron/8.0 mil	6061-F	210	18
B	Boron/8.0 mil	6061-T6	230	37
C	Boron/5.6 mil	6061-F	180	18
C	Boron/5.6 mil	6061-T6	175	25

Table 9.^a Properties of 50 v/o Borsic/titanium

Source	Filament diameter	matrix	Longitudinal tensile strength (ksi)	Transverse (ksi) tensile strength
A	4.2 mil	Beta III	170	30
A	4.2 mil	6A1-4V	155	35
B	4.2 mil	6A1-4V	140	42
C	5.7 mil	6A1-4V	185	64

^aExtracted from C. T. Lynch and J. P. Kershaw, Metal Matrix Composites, Chemical Rubber Company (CRC), Boca Raton, FL c1972. Permission granted by CRC.

Table 10. Axial tensile strength of 5.6 mil B/A1

Matrix	v/o Boron (%)	Ultimate tensile strength (10 ³ psi)	Elastic modulus (10 ⁴ psi)	Strain to fracture (%)
2024F	45	185.7	30.4	0.765
	45	197.5	27.5	0.835
	44	177.0	30.0	0.725
	47	212.0	32.0	0.825
	47	212.0	32.6	0.820
	49	194.0	32.0	0.740
2024-T6	46	202.5	32.8	0.75
	46	213.6	31.6	0.81
	47	217.0	32.3	0.830
	48	213.0	31.3	0.845
	64	279.0	40.0	0.755
2024F	70	279.5	—	—
	66	253.0	—	—
	67	250.2	—	—
6061F	48	196.3	31.8	0.710
	48	171.0	28.2	0.590
	50	204.0	33.8	0.72
	50	208.0	32.0	0.76
6061-T6	52	216.5	33.8	0.78
	51	197.0	33.4	0.69
	50	203.0	—	—

Table II. Axial tensile strength of 5.7 mil Borsic/Al composites

Matrix	v/o BORSIC (%)	Ultimate tensile strength (10 ³ psi)	Elastic modulus (10 ⁴ psi)	Strain to fracture (%)
6061-F	30	115.0	17.6	0.71
		113.3	18.9	0.71
6061-T6	30	156.2		
		152.4		
6061-F	54	203.4	36.6	0.675
		181.5	36.0	0.630
		199.0	36.1	0.655
6061-F	56	214.0		
		212.0		
6061-F	57	228.0		
		222.0		
6061-F	58	227.0		
		219.0		
		216.0		
		222.0		
6061-F	61	199.0	39.4	0.57
		207.6		
2024-F	58	211.5		
		221.0		
2024-T6	61	235.0		
		210.0		
5052/56	59	177.6	37.7	0.54
		182.0		
1100/1145	57	158.2		
		175.5		

Table 12. Transverse tensile properties of 5.6 mil B/Al

Matrix	v/o Boron	UTS (10 ³ psi)	Elastic modulus (10 ⁶ psi)	Strain to failure (%)
2024F	45	27.0		
	45	27.2		
	45	26.2		
2024-T6	45	48.0		
	45	48.7		
	45	38.2		
2024-T6	55	41.9	21.0	0.23
		45.0	22.5	0.24
2024F	66	26.0		
		27.3		
6061F	50	18.9		
	50	19.0		
	50	18.3		
6061-T6	50	34.8		
	50	37.4		
	50	41.7		

Table 13. Transverse tensile properties of 5.7 mil Borsic/Al

Matrix	Boron v/o	UTS (10 ³ psi)	Elastic modulus (10 ⁴ psi)	Strain to failure (%)
6061-F	30	16.0 14.0	15.0	0.45
6061-T6	30	35.1 34.6	15.5	0.31
6061-F	52	19.3 20.5 19.3		
6061-T6	52	31.2 36.6 39.8	18.9 19.0 21.0	0.19 0.23 0.24
6061-F	60	19.9 19.7	24.0	0.26
6061-T6	60	44.0 37.2		0.23
6061-F	70	16.8 21.2 18.8	32.3	0.29
6061-T6	70	35.4 27.6	29.2	0.14
2024-F	56	22.4 20.8	22.8	0.171
2024-T6	56	46.0 37.1	24.7	0.202
5052/56	57	30.4 32.6	25.8	0.308 0.50
1100	54	13.6 13.8 11.7	21.8 23.2	0.68 0.52 0.60

Table 14. Axial tensile strength of
48 v/o 5.6 mil B/6061 Al composites.

Test temperature (°F)	UTS (10 ³ psi)
70	196.3
70	171.0
500	177.0
500	135.2
900	143.8
900	137.3

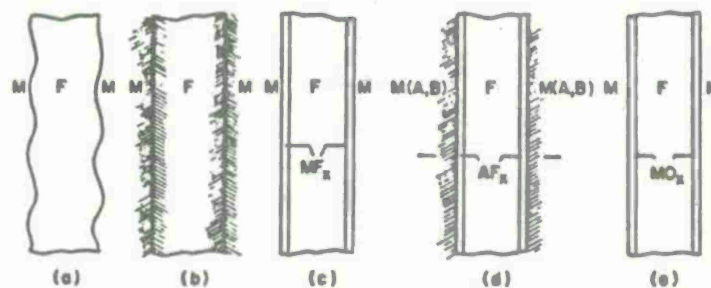


FIG. 1.^a Schematic of principal bond types: matrix M may contain elements A and B and filament may be single element (such as graphite) identified as F or compound (such as Al_2O_3) denoted as FO_x . (a) Mechanical bond, (b) dissolution and wetting bond, e.g., Cb-W, (c) reaction bond e.g., Ti-C, (d) exchange reaction (e.g., Ti(Al)-B), (e) pseudo-Class I oxide bond where $\Delta F_{MO_x} < \Delta F_{FO_x}$ (e.g., Al-B).

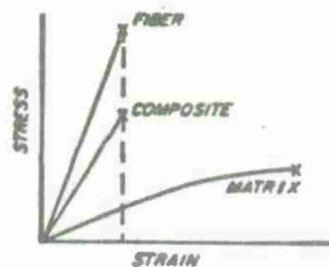


Figure 2.^b Typical stress-strain relations of fiber, matrix and composite. The composite failure strain is close to the fiber failure strain. The matrix is nonlinear above the fiber failure strain.

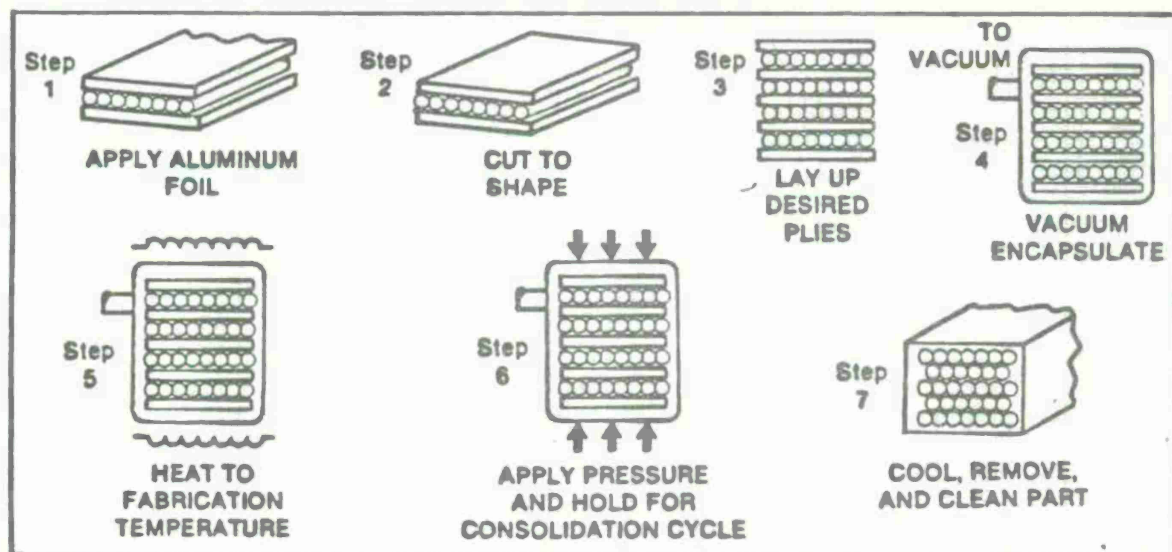


Figure 3. Composite materials fabrication process.

^aExtracted from A. G. Metcalfe, "Interfaces in Metal Matrix Composites," Composite Materials, Vol. I, Academic Press, New York, New York, c1974. Permission granted by Academic Press.

^bExtracted from S. W. Tsai and H. T. Hahn, Introduction to Composite Materials, Technomic Publishing Company, Lancaster, PA, c1980. Permission granted by Technomic Publishing Company.

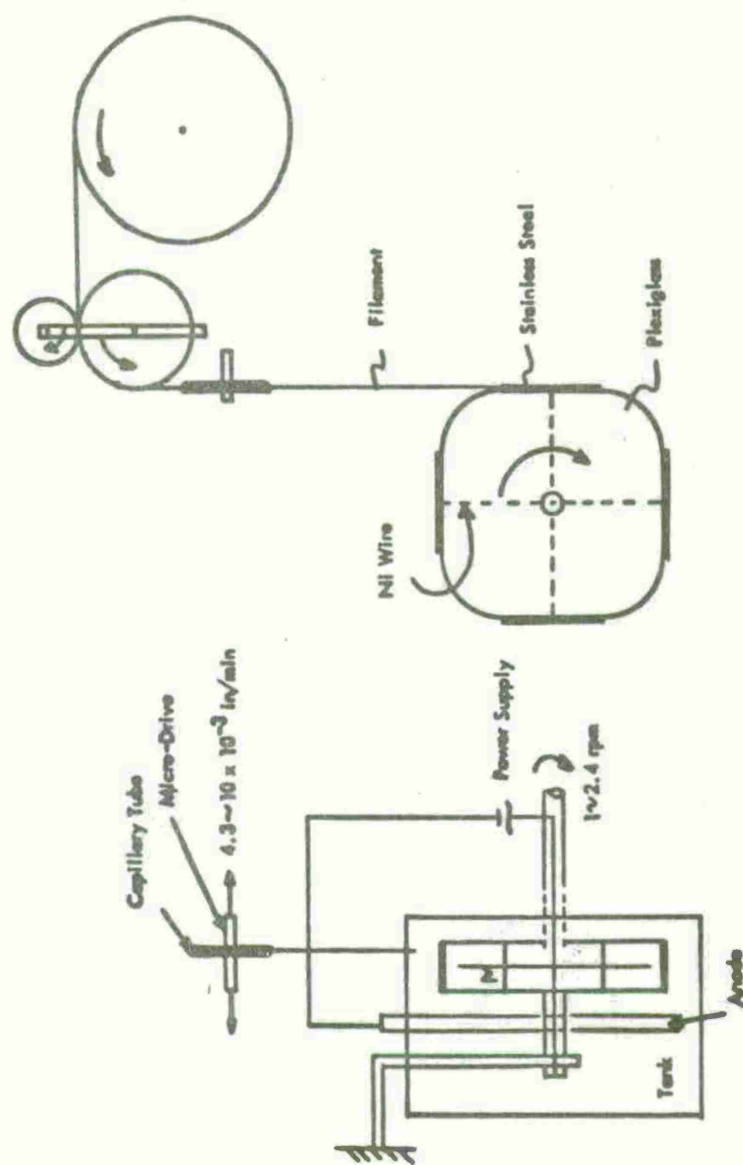


Figure 4. A schematic diagram of the winding apparatus.

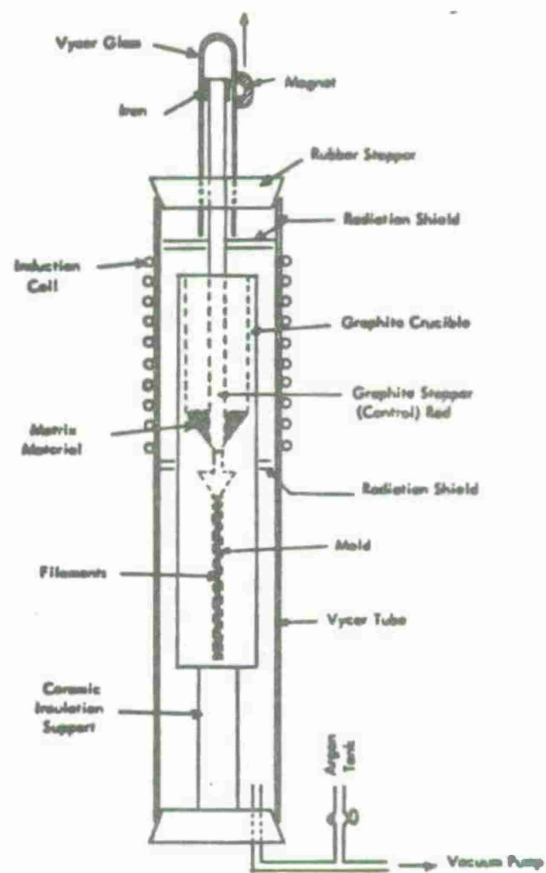


Figure 5. Schematic diagram of the apparatus for vacuum casting.

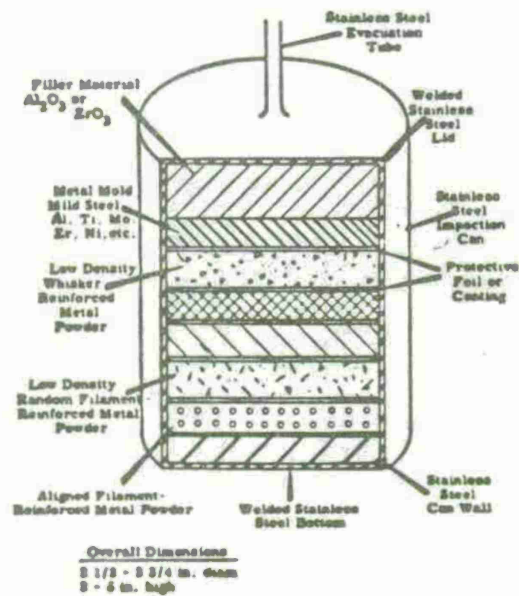


Figure 6. Composite billet for high energy-rate forming composite plates.

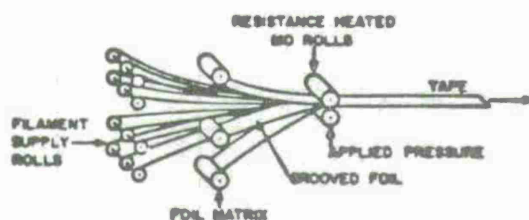


Figure 7. Hot-roll bonding.

$$F_{xs} = F_{ys} = F_s = 0$$

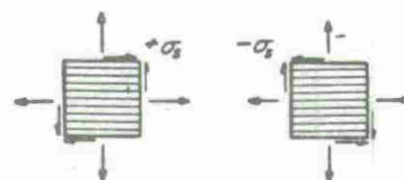


Figure 8. On-axis positive and negative shears. They should have no effect on the strength of unidirectional composites. Coupling between shear and normal components cannot exist in this orthotropic orientation.

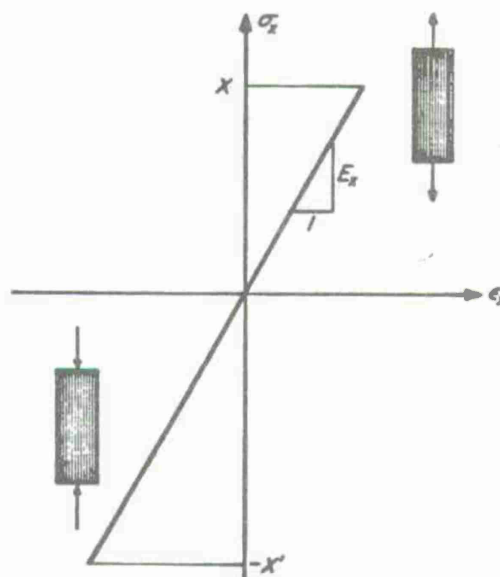


Figure 9. Uniaxial longitudinal tensile and compressive tests.

^a Extracted from S. W. Tsai and H. T. Hahn, Introduction to Composite Materials, Technomic Publishing Company, Lancaster, PA, c1980. Permission granted by Technomic Publishing Company.

^b Extracted from A. G. Metcalfe, "Interfaces in Metal Matrix Composites," Composite Materials, Vol I, Academic Press, New York, New York, c1974. Permission granted by Academic Press.

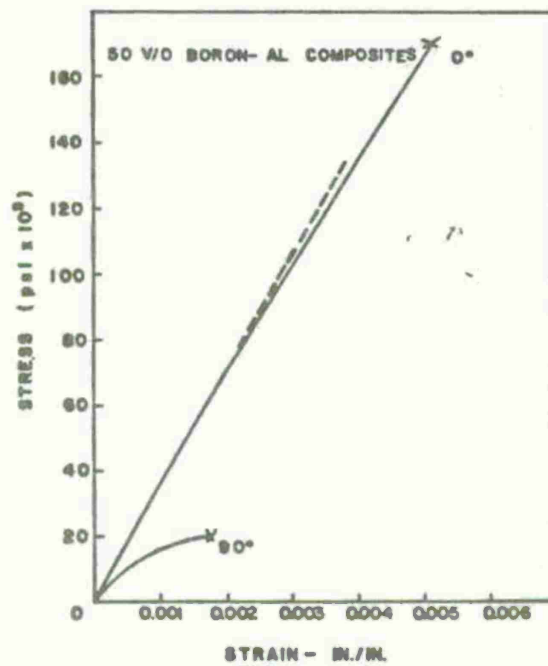


Figure 10. Typical stress-strain curves for 50 vol % unidirectional boron/aluminum composites tested parallel (0°) and perpendicular (90°) to the filament.

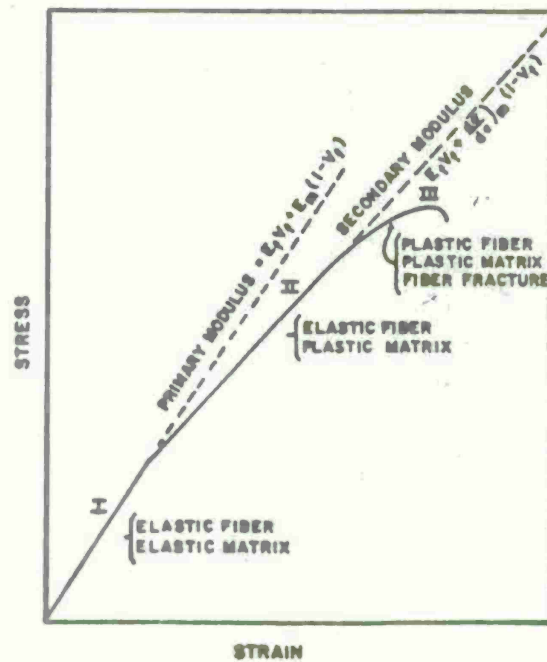


Figure 11. Schematic stress-strain curve for filamentary reinforced metals showing three regions.

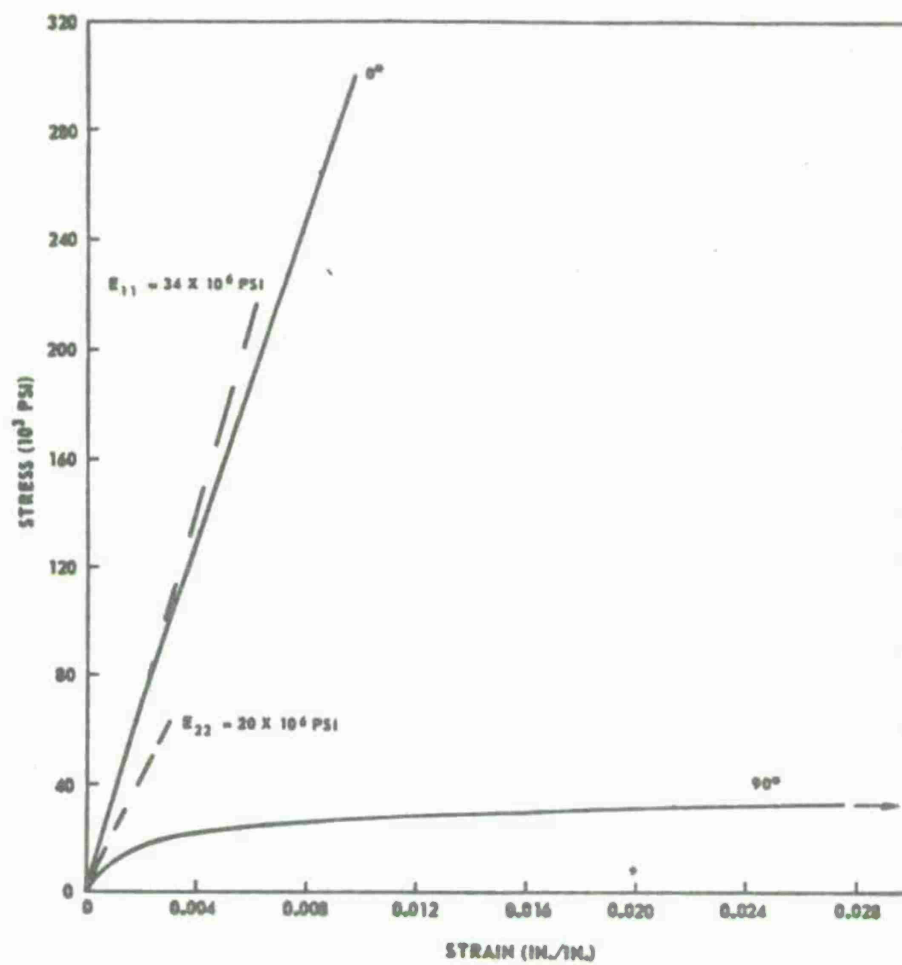


Figure 12. Typical compressive stress-strain curves for 50 vol % unidirectional boron/aluminum composites tested parallel (0°) and perpendicular (90°) to the filament.

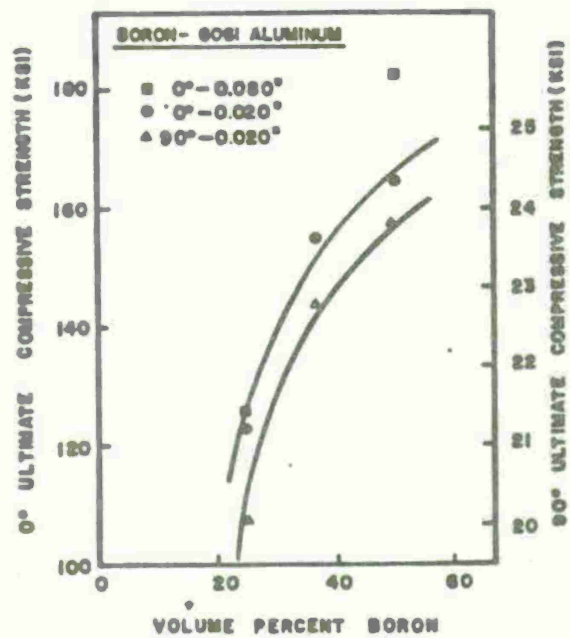


Figure 13. Variation of longitudinal and transverse compressive strength with boron content for unidirectional boron/aluminum composites.

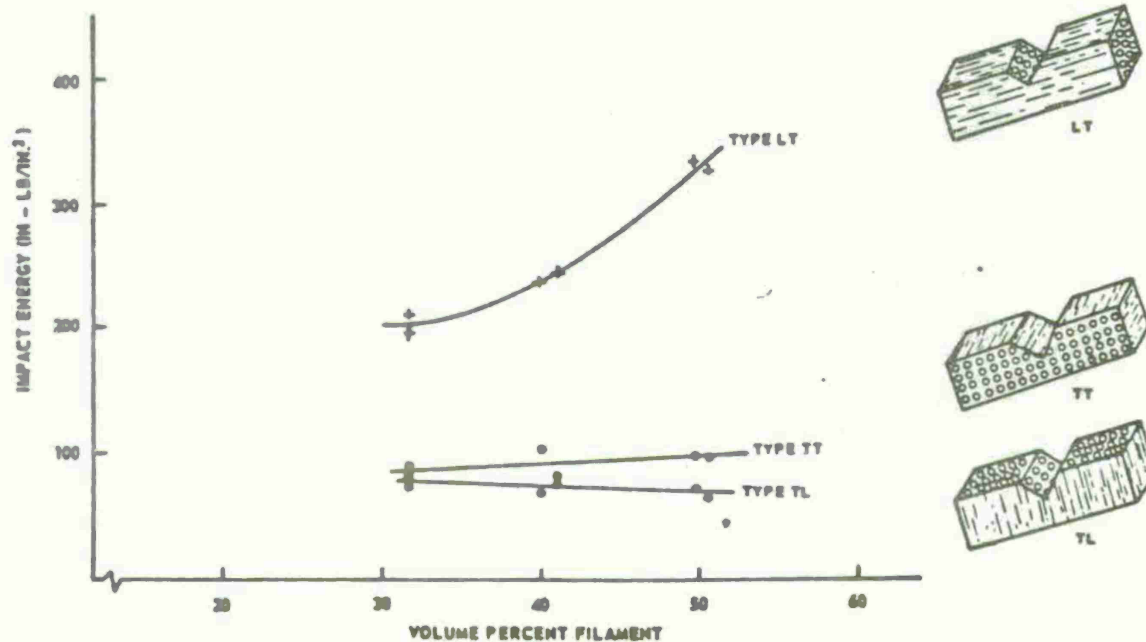


Figure 14. Variation of impact energy with filament content for various notch-filament configurations for 46 vol % Borsic-6061 aluminum unidirectional composites.

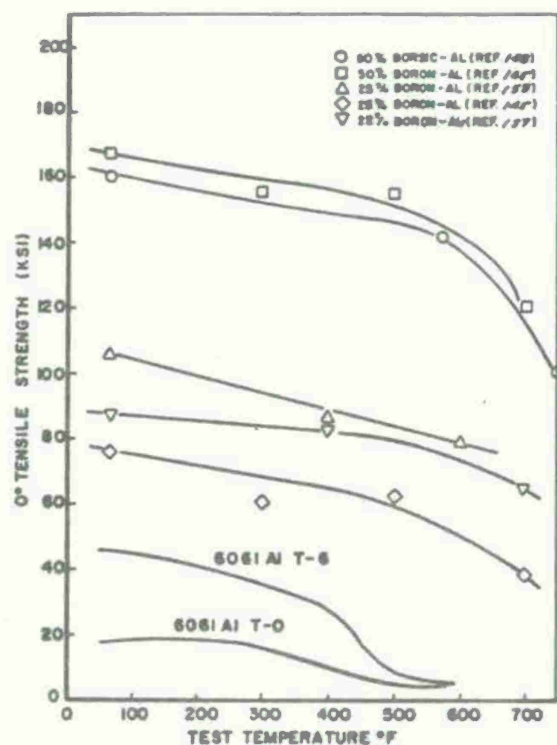


Figure 15. Variation of longitudinal (0°) tensile strength with test temperature for unidirectional boron (Borsic)-aluminum composites.

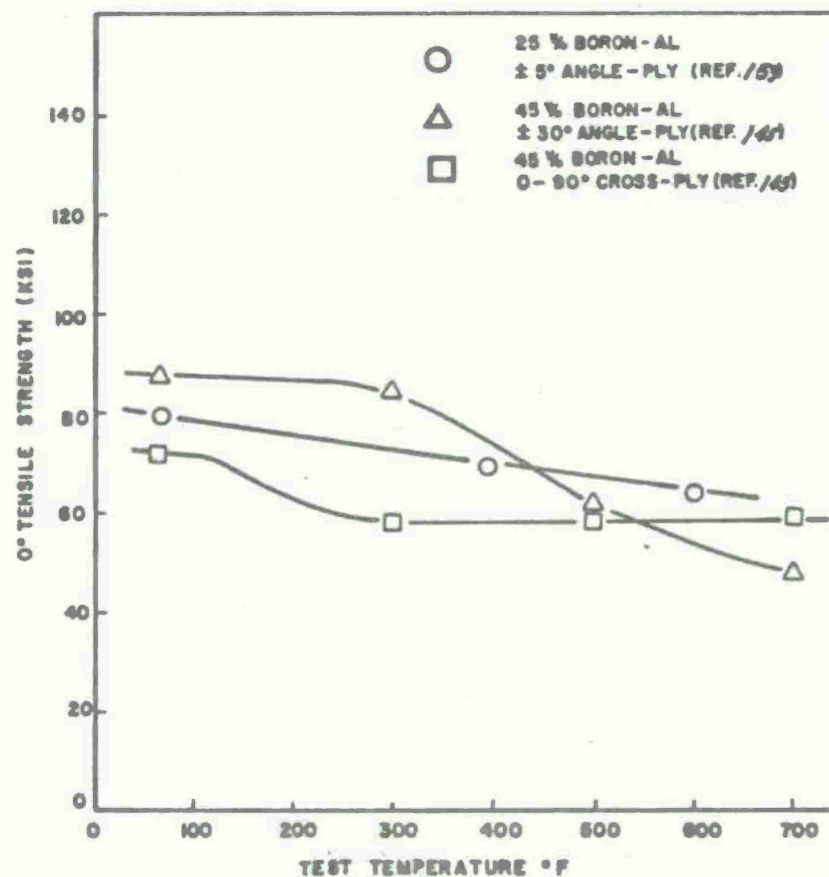


Figure 16. Variation of longitudinal (0°) tensile strength with test temperature for various cross- and angle-ply boron-aluminum composites.

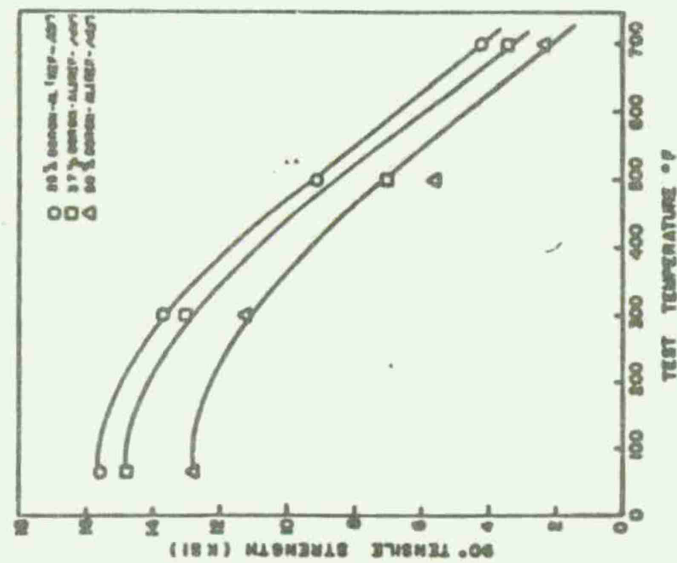


Figure 17. Variation of transverse (90°) tensile strength with test temperature for unidirectional boron-aluminum composites.

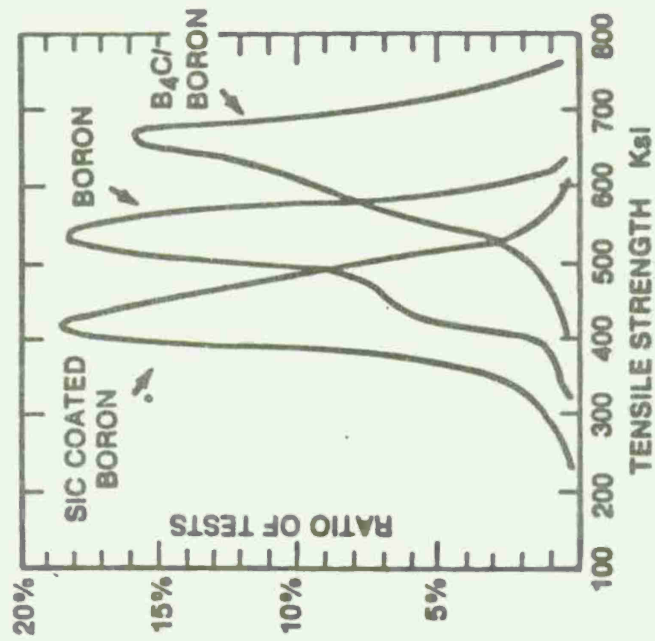


Figure 18. A typical histogram for the strength of boron, SIC coated boron and B₄C/boron filament.

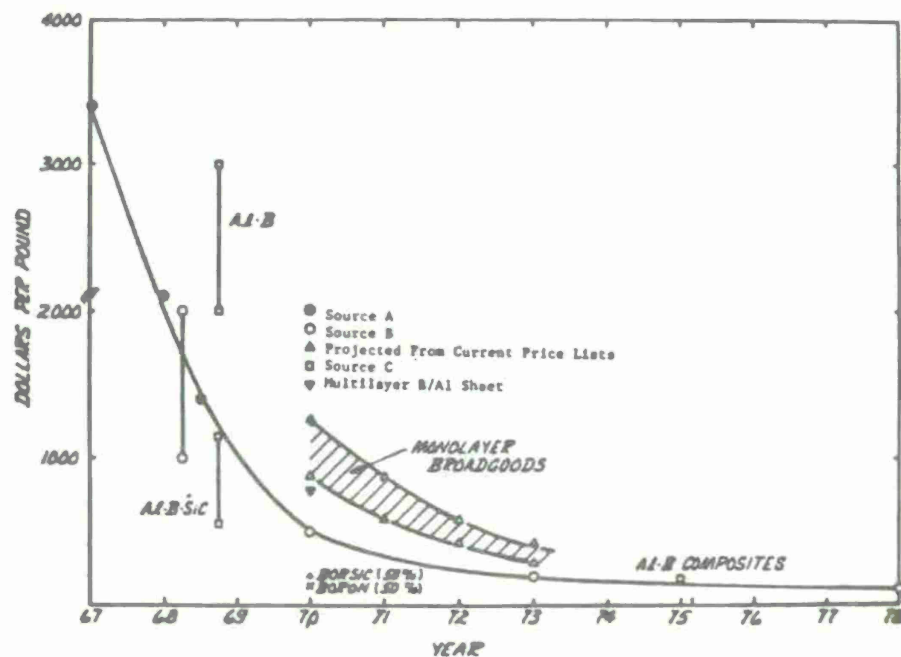


Figure 19. Metal matrix composite cost history and projections.

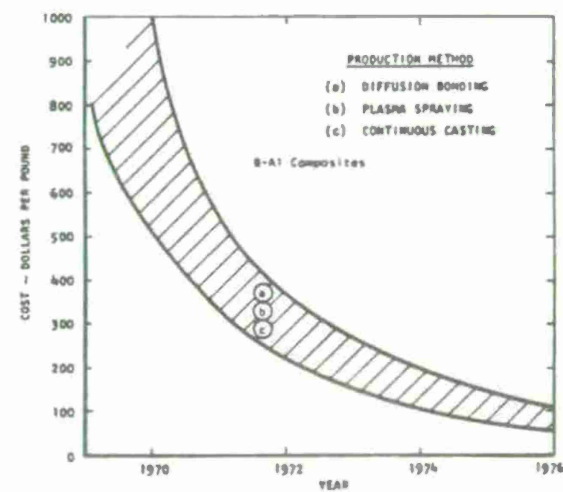
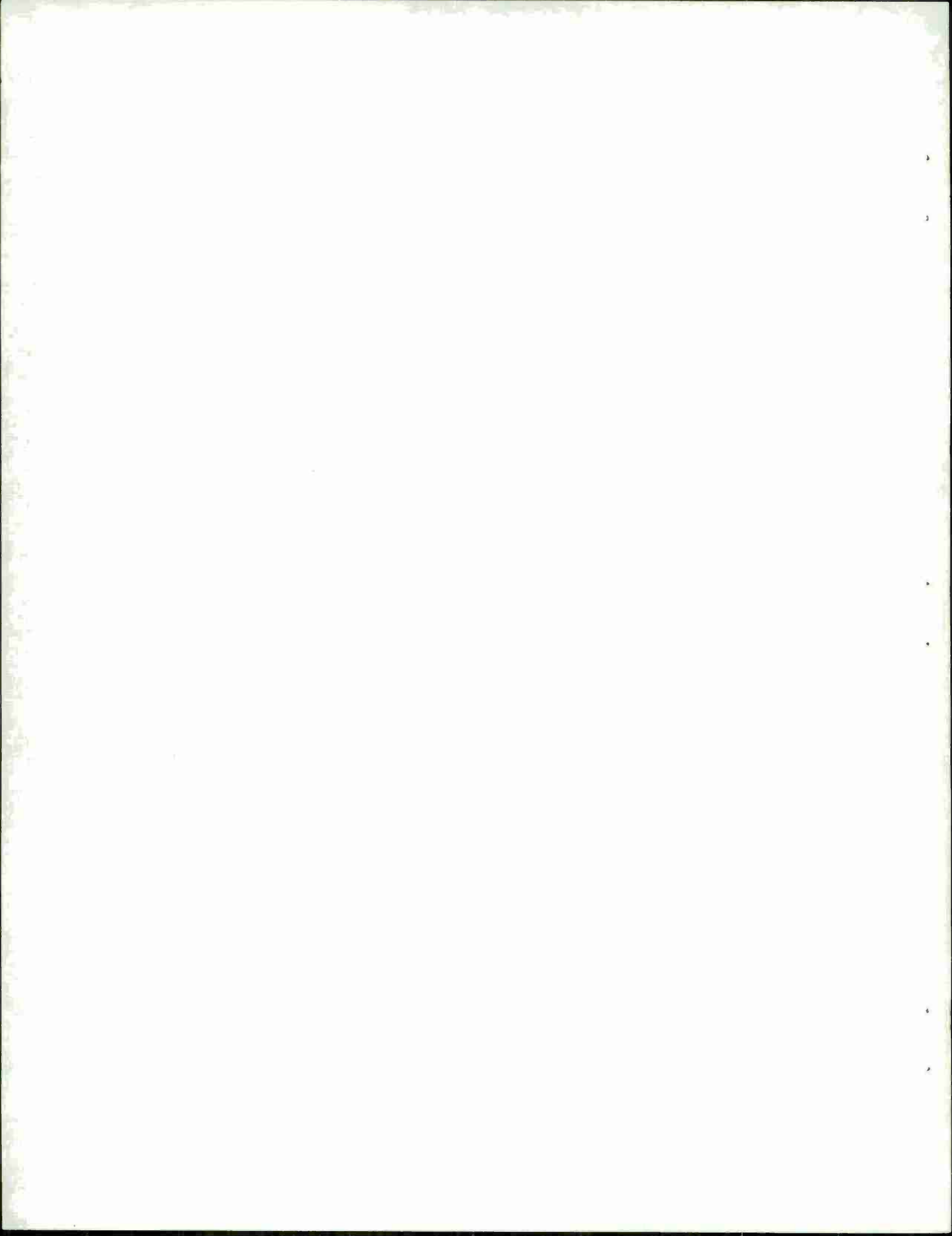


Figure 20. Current and projected costs for fibers and composites.



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